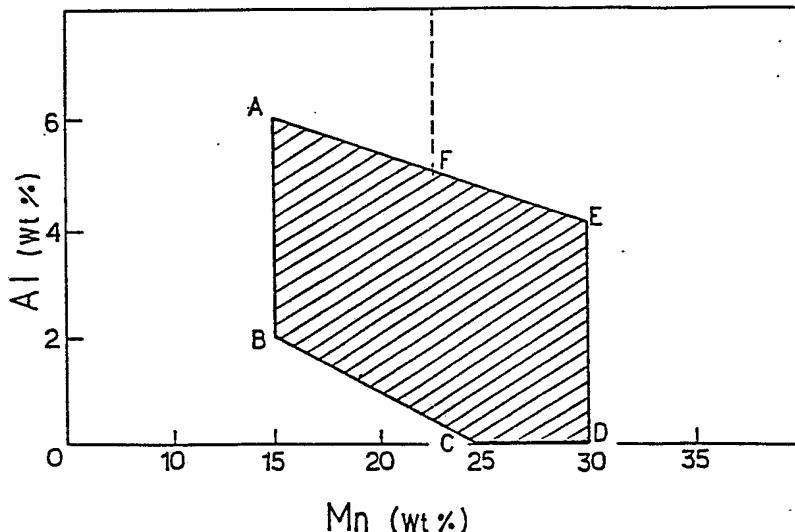




INTERNATIONAL APPLICATION PUBLISHED UNDER THE PATENT COOPERATION TREATY (PCT)

(51) International Patent Classification 5 : C22C 38/06, 38/08, 38/12 C22C 38/14, 38/16, 38/18 C21D 7/13, 7/02		A1	(11) International Publication Number: WO 93/13233 (43) International Publication Date: 8 July 1993 (08.07.93)
(21) International Application Number: PCT/KR92/00082		(74) Agent: CHUN, Choon, Hang; 3rd Floor, Kyongbok Bldg., 277-27, Nonhyun-Dong, Kangnam-ku, Seoul 135-010 (KR).	
(22) International Filing Date: 29 December 1992 (29.12.92)		(81) Designated States: BR, CA, JP, RU, US, European patent (DE, ES, FR, GB, IT).	
(30) Priority data: 91-25112 30 December 1991 (30.12.91) KR 92-13309 24 July 1992 (24.07.92) KR		Published <i>With international search report.</i>	
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(54) Title: AUSTENITIC HIGH MANGANESE STEEL HAVING SUPERIOR FORMABILITY, STRENGTH AND WELDABILITY, AND MANUFACTURING PROCESS THEREFOR



(57) Abstract

An austenitic high manganese steel having superior formability, strength and weldability, and a process for manufacturing the steel, are disclosed. The superior formability of the steel is suitable for use on automobiles and electronic panel. The steel has a composition of (in weight %) less than 1.5 % of C, 15.0-35.0 % of Mn, 0.1-6.0 % of Al, and the balance of Fe and other indispensable impurities. The size of the austenite grains is less than 40.0 μm , and, one or more elements are added by selecting them from a group consisting of less than 0.60 % of Si, less than 5.0 % of Cu, less than 1.0 % of Nb, less than 0.5 % of V, less than 0.5 % of Ti, less than 9.0 % of Cr, less than 4.0 % of Ni, and less than 0.2 % of N, thereby providing an austenitic high manganese steel having superior formability, strength and weldability.

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AUSTENITIC HIGH MANGANESE STEEL HAVING SUPERIOR
FORMABILITY, STRENGTH AND WELDABILITY, AND
MANUFACTURING PROCESS THEREFOR

5 Field of the invention

The present invention relates to an austenitic high manganese steel which is used in fields requiring a high formability such as automobile steel sheet, electronic panel sheet, and the like. Particularly the present 10 invention relates to an austenitic high manganese steel having a good formability, high strengths and superior weldability.

Background of the invention

15 In the application field of steel, those which require best formability are automobile steel sheets, and electronic panel sheets.

Particularly, in the automobile industry, the discharge of carbon dioxide is more strictly regulated 20 coming recently for alleviating the air pollution. In accordance with this trend, there has been demanded a high strength steel sheet which has a good formability, as well as improving the combustion rate of the fuel, and reducing the weight of the automobile.

25 Conventionally, as the automobile steel sheet, a extra low carbon steel in which the matrix structure is a ferrite has been used for assuring the formability (U.S. Patents 4,950,025, 4,830,686 and 5,078,809).

However, in the case where the extra low carbon steel 30 is used for the automobile steel sheet, although the formability is superior, the tensile strength is lowered

to 28-38 kg/mm². Consequently the weight of the automobile cannot be reduced, and the safety of the automobile is lowered, thereby jeopardizing the lives of passengers.

5 The extra low carbon steel having the ferrite matrix can include up to 0.005 % of carbon, and the solubility limit for impurities is very low. If carbon and other impurities are added in excess of the solubility limit, then carbides and oxides are formed, with the
10 result that particular textures cannot be developed during cold rolling and annealing processes, thereby degrading the formability.

Thus, in the case of the conventional automobile steel sheet having the ferrite matrix, the addition of carbon is
15 reduced to about 0.003%, as well as reducing other impurities to extremely small amounts for enhancing the formability. Consequently, there are accompanied difficulties such that special treatment such as degassing treatment has to be carried out in the steel making process, and that particular textures have to be developed
20 during cold rolling and annealing processes.

Further, a multi-phase steel in which the low strengths of the extra low carbon steel are improved is disclosed in U.S. Patent 4,854,976. In this steel, Si,
25 Mn, P, Al and B are added in large amounts to form a bainite structure and retained austenite structure of less than 8%, thereby increasing the tensile strength to 50-70 kg/mm². However, due to the difference of the deformation capabilities between the bainite structure and the retained austenite structure, the formability is lowered, and therefore, this material is limitedly used
30

in automobile parts which do not require a high formability.

Meanwhile, the steel sheet which is used as the external panel of electronic apparatus has to be non-magnetic material which is not influenced by magnetic fields, as well as being high in its strengths and formability. Therefore, austenitic stainless steel is mainly used for this purpose, but this steel contains expensive nickel to about 8%, while its magnetic susceptibility becomes unstable due to strain-induced α' -martensites during its manufacturing process.

The present inventors have been engaged for many years in studying on how to overcome the disadvantages of the conventional automobile steel sheet and the electronic steel sheet, and have successfully developed an austenitic high manganese steel having superior formability and strengths.

So far, no case has been found in which a high manganese steel is used to attempt providing good formability and high strength.

Currently,, the high manganese steel is used in nuclear fusion reactor, in magnetic floating rail for the purpose of preventing electrostatic charges, and as non-magnetic structural material for transformers (Japanese Patent Laying-opening No. Sho-63-35758, 64-17819, 61-288052 and 60-36647). Further, this material is also used as non-magnetic steel for some parts of VTR and electronic audio apparatuses (Japanese Patent Laying-opening No. Sho-62-136557).

However, in this non-magnetic high manganese steel, either Al as an ingredient of the alloy is not added, or

it is added up to only 4% for deoxidizing, oxidation resistance, corrosion resistance, solid solution hardening, and grain refinement (Japanese Patent Laying-opening No. Sho-60-36647, 63-35758, and 62-136557)

5 Meanwhile the alloy of the same composition system which is related to the present invention is disclosed in Korean Patent 29304 (the corresponding U.S. Patent 4,847,046, and Japanese Patent 1,631,935) which is granted to the present inventors).

10 However, the alloy system which is disclosed in Korean Patent 29304 is considered on its ultra low temperature strength and toughness, and therefore, is for being used in the cryogenic applications. Therefore, it is essentially different from the steel of the present 15 invention which is intended to improve the formability, strengths and weldability.

Summary of the invention

Therefore, it is an object of the present invention 20 to provide an austenitic high manganese steel and a manufacturing process thereof, in which the fact that an austenitic Fe-Mn-Al-C steel having a face centered cubic lattice has a high elongation is utilized to produce a proper amount of strain twins, thereby improving the 25 formability, strengths and weldability.

It is another object of the present invention to provide an austenitic high manganese steel and a process for preparation thereof, in which a solid solution hardening element is added into an austenitic Fe-Mn-Al-C having a face centered cubic lattice, so that the strain 30 twins should further improve the formability, strength and

weldability.

Brief description of the drawings

5 The above object and other advantages of the present invention will become more apparent by describing in detail the preferred embodiment of the present invention with reference to the attached drawings in which:

Figure 1 is a graphical illustration showing the addition ranges of Mn and Al;

10 Figure 2 is a graphical illustration showing the limits of the formability based on the experiments;

Figure 3 is an electron micrograph showing the formation of strain twins in the steel of the present invention;

15 Figure 4 is an electron micrograph showing the formation of deformation twins in another embodiment of the present invention;

Figure 5 is a graphical illustration showing the limit of the formability based on the experiments; and

20 Figure 6 is a graphical illustration showing the variation of a hardness on the welded joint based on the experiments.

Description of the preferred embodiment

25 The steel of the present invention contains less than 0.70 weight % of C, and Mn and Al are added so as to come within the range which is enclosed by A, B, C, D and E in Figure 1. The remaining part consists of Fe and other indispensable impurities, thereby forming an austenitic 30 high manganese steel which has superior formability, strengths and weldability.

After a long study and experiments, the present inventors found that, even if the C, Mn and Al of the austenitic high manganese steel is varied to a certain degree, and even if the solid solution hardening element is added, still a high manganese steel having superior formability, strengths and weldability can be obtained.

Based on this fact, a new invention is embodied, and this new invention will be described in detail below.

The steel of the present invention is composed of in weight % less than 1.5% of C, 15.0-35.0% of Mn, and 0.1-6.0% of Al, the balance consisting of Fe and other indispensable impurities. The grain size is 40.0 μm , and the formability, strengths and weldability are superior.

In another embodiment, the steel of the present invention is composed of in weight % less than 1.5% of C, 15.0-35.0% of Mn, 0.1-6.0% of Al, and one or more selected from the group consisting of less than 0.60% of Si, less than 5.0% of Cu, less than 1.0% of Nb, less than 0.5% of V, less than 0.5% of Ti, less than 9.0% of Cr, less than 4.0% of Ni, and less than 0.12% of N. The balance includes Fe and other indispensable impurities while the grain size is smaller than 40.0 μm , thereby providing an austenitic high manganese steel having superior formability, strength and weldability.

The high manganese steel of the present invention is hot-rolled and cold-rolled sequentially.

The manufacturing process of the steel of the present invention consists of such that a steel slab containing in weight % less than 1.5% of C, 15.0-35.0% of Mn, 0.1-6.0% of Al, and the balance of Fe and other indispensable impurities is prepared, and the steel slab is hot-rolled

to hot rolled steel sheet in the normal method. Or the hot rolled steel sheet is cold rolled, and then, it is annealed at a temperature of 500-1000°C for 5 seconds to 20 hours, thereby obtaining an austenitic high manganese steel having superior formability, strengths and weldability.

Alternatively, the manufacturing process of the steel of the present invention consists of such that a steel slab is prepared, the slab containing in weight % less than 1.5 of C, 15.0-35.0 of Mn, 0.1-6.0 of Al, and one or more elements selected from the group consisting of less than 0.60% of Si, less than 5.0% of Cu, less than 1.0% of Nb, less than 0.5% of V, less than 0.5% of Ti, less than 9.0% of Cr, less than 4.0% of Ni, and less than 0.2% of N. The balance consists of Fe and other indispensable impurities, and this slab is hot-rolled to hot rolled steel sheet as the final product. Or alternatively the hot rolled steel sheet is cold-rolled, and then, it is annealed at a temperature of 550-1000°C for 5 seconds to 20 hours, thereby obtaining an austenitic high manganese steel having superior formability, strengths and weldability.

Now the reason for the selection of the alloying elements and the addition ranges will be described.

The carbon (C) inhibits the formation of ϵ -martensites by increasing the stacking fault energy, and improves the stability of the austenite. However, if its content is over than 1.5 weight % (to be called %), its stacking fault energy becomes too high, with the result that no twins can be formed. Further, the solubility limit of carbin in the austenite is exceeded, with the result that

carbides are excessively precipitated, thereby deteriorating the elongation and formability. Thus the content of carbon should be desirably less than 1.5%.

The manganese (Mn) is an indispensable element for improving the strengths and for stabilizing the austenite phase. However, if its content is less than 15.0%, an α' -martensite phase come to exist, while if its content is over 35.0%, the formation of twins is inhibited because its addition effect is annulled. Therefore the content of manganese should be desirably confined within 15.0-35.0%.

The aluminum (Al) like the carbon heightens the stacking fault energy to stabilize the austenite phase, and does not form ϵ -martensites even under a severe deformation such as cold rolling, but contributes to forming twins. Thus the aluminum is an important element for improving the cold workability and press formability. However, if its content is less than 0.1%, ϵ -martensites are formed to deteriorate the elongation, although its strengths are reinforced, with the result that cold workability and press formability are deteriorated. Meanwhile, if its content exceeds 6.0%, the stacking fault energy is too much augmented, so that a slip deformation occurs due to a perfect dislocation. Therefore, the content of aluminum should be desirably 0.1-6.0%.

As described above, the addition of manganese and aluminum inhibits the formation of α' -martensites, and excludes the possibility of the formation of ϵ -martensites and slip deformations due to a perfect dislocation. Thus the two elements are limited so as for twins to be formed

owing to partial dislocations.

The Si is an element added to deoxidize and to improve strengths by solution-hardening. If its content is over 0.6%, the deoxidizing effect is saturated, and the paint 5 coatability is deteriorated during the manufacturing of cars, while cracks are formed during welding. Therefore the content of Si should be desirably limited to below 0.60%.

The Cu is an element to be added for the improvement 10 of corrosion resistance and the increase of strengths through a solid solution hardening. If its content is over 5.0%, a hot brittleness occurs so as for hot rolling to be impaired. Therefore the content of Cu should be desirably limited to below 5.0%.

15 The Nb, V and Ti are elements to be added for improving strengths through a solid solution hardening. If the content of Nb is over 1.0%, cracks are formed during hot rolling, while if the content of V is over 0.5%, low melting point chemical compounds are formed, 20 thereby impairing hot rolling quality. Meanwhile, the Ti reacts with nitrogen within the steel to precipitate nitrides, and consequently, twins are formed, thereby improving strengths and formability. However, if its content is over 0.5%, excessive precipitates are formed, 25 so that small cracks should be formed during cold rolling, as well as aggravating formability and weldability. Therefore, the contents of Nb, V and Ti should be limited to respectively 1.0%, 0.5% and 0.5%.

30 The Cr and Ni are elements to be added for inhibiting the formation of α' -martensite by stabilizing the austenite phase, and for improving strengths through a solid

solution hardening. If the content of Cr is less than 9.0%, the austenite phase is stabilized, and prevents the formation of cracks during the heating of slab and during hot rolling, thereby improving the hot rollability.

5 However, if its content is over 9.0%, α' -martensites are produced in large amounts, thereby deteriorating the formability. Therefore, the content of Cr should be desirably limited to below 9.0%. The Ni improves elongation, and also improves mechanical properties such as impact strength. However, if its content exceeds 4.0%, its addition effect is saturated, and therefore, its content should be desirably limited to 4.0% by taking into account the economic aspect.

The nitrogen (N) precipitates nitrides in reaction with Al in the solidification stage, during the hot rolling stage, and during the annealing stage after the cold rolling, and thus, performs a core role in producing twins during the press forming of steel sheets, thereby improving the formability and strengths. However, if its content exceeds 0.2%, the nitrides are precipitated in an excessive amount, thereby aggravating the elongation and the weldability. Therefore, the content of N should be desirably limited to below 0.2%.

Now the present invention will be described as to its manufacturing conditions.

The steel which has the above described composition undergoes a number of processes such as melting, continuous casting (or ingot casting) and hot rolling. As a result, a hot rolled steel plate having a thickness of 1.5-8 mm are obtained to be used on trucks, buses and other large vehicles.

This hot rolled steel sheet is cold-rolled and annealed into a cold rolled sheet of below 1.5 mm to be used mainly for motor vehicles. As to the annealing heat treatment, either continuous annealing heat treatment or 5 box annealing heat treatment is possible. However, the continuous annealing heat treatment is preferable because of its economical feature in mass production.

The hot rolling for the steel of the present invention is carried out in the normal manner, and preferably, the 10 slab reheating temperature should be 1100-1250°C, while the finish hot rolling temperature should be 700-1000°C. The above mentioned hot rolling temperature of 1100-1250°C is adopted so that the slab should be uniformly heated within a short period of time in order to improve the 15 energy efficiency. If the hot rolling finish temperature is too low, the productivity is diminished, and therefore, its lower limit should be 700°C. The upper limit of the hot rolling finish temperature should be 1000°C, because over 10 rolling passes have to be 20 undergone during the hot rolling process.

The cold rolling is also carried out in the normal manner. In manufacturing the Fe-Mn-Al-C steel, if the annealing temperature is below 500°C, then deformed austentic grains cannot be sufficiently recrystallized. 25 Further, in this case, rolled elongated grains remain, and therefore, the elongation becomes too low, although the strengths are high. Meanwhile, if the annealing temperature is over 1000°C, austenite grains are grown into over 40.0 μm, with the result that the formability 30 is lowered. Therefore the annealing temperature should be preferably limited to 500-1000°C.

If the annealing time is less than 5.0 seconds, the heat cannot reach to the inner portion of the cold rolled sheet, with the result that complete recrystallizations cannot be formed. Further, in this case, the cold 5 rolled grains remain, so that the formability should be impaired. Meanwhile, if the annealing time exceeds 20 hours, the time limit is violated to form coarse carbides, thereby lowering the strengths and the formability. Therefore the annealing time should be preferably limited 10 to 5 seconds to 20 hours.

In the case where the Fe-Mn-Al-C steel is manufactured by adding a solid solution hardening element, it is desirable to limit the annealing temperature and the annealing time to 550-1000°C and to 5.0 seconds to 20 hours 15 respectively for the same reason described above.

The hot rolled steel sheet which is manufactured through the stages of alloy design - melting - continuous casting -hot rolling according to the present invention is cold rolled and annealed, so that the size of the austenite grains should be less than 40 μm , the tensile strength should be over 50 kg/mm², and the elongation should be over 40%. 20

In the steel of the present invention, if the grain size is over 40 μm , the formability is aggravated, and therefore, an adjustment for the annealing should be made 25 in order to reduce the grain size to be smaller than 40 μm .

Now the present invention will be described further in detail based on actual examples.

30 <Example 1>

A steel having the composition of Table 1 below was

melted in vacuum, and then, steel ingots of 30 kg were formed. Then a solution treatment was carried out, and then, a slab rolling was carried out to form slabs having a thickness of 25 mm.

5 The slab manufactured in the above described manner was heated to a temperature of 1200°C, and a hot rolling was carried out, with the finish rolling temperature being 900°C. A hot rolled plate of a thickness of 2.5 mm was produced by this hot rolling process, and then, this
10 hot rolled plate was cold rolled into a thickness of 0.8 mm.

The cold rolled sheet was annealed at a temperature of 1000°C for 15 minutes, and an X-ray diffraction test was carried out on each of the test pieces. Then the
15 volume fraction of the phases at the room temperature was observed, and this is shown in Table 1 below. Further, the permeability of the each of the test pieces was measured, this being shown also in Table 1 below.

Further, tensile tests were carried out on the test
20 pieces for tensile strength, yield strength and elongation. Further, the uniformloy elongated portion of the tensile specimen after the tensile tests was cut out, and an X-ray diffraction test was carried out on the portion to measure volume fractions of strain-induced phase, this data being
25 shown in Table 2 below.

Table 1

Steel type	Chemical composition (weight%)								Volume fractions of the phases			Peameabi- lity (H=10000e)
	C	Mn	P	S	Al	Ti	Cr	Ni	γ (auste- nite)	ε marten- site	α' marten- site	
Steel of the invention	1	0.64	15.5	-	-	3.0	-	-	100	-	-	1.0003
	2	0.38	17.9	-	-	3.3	-	-	100	-	-	1.0003
	3	0.27	19.1	-	-	3.2	-	-	100	-	-	1.0003
	4	0.36	19.1	-	-	3.6	-	-	100	-	-	1.0003
	5	0.13	22.7	-	-	1.9	-	-	100	-	-	1.0003
	6	0.13	23.0	-	-	4.0	-	-	100	-	-	1.0003
	7	0.47	23.1	-	-	3.5	-	-	100	-	-	1.0003
	8	0.07	23.8	-	-	1.1	-	-	100	-	-	1.0003
	9	0.34	24.8	-	-	1.3	-	-	100	-	-	1.0003
	10	0.13	25.3	-	-	0.3	-	-	100	-	-	1.0003
	11	0.12	27.2	-	-	3.1	-	-	100	-	-	1.0003
	12	0.43	28.7	-	-	0.5	-	-	100	-	-	1.0003
Comparative steel	13	0.06	14.4	-	-	2.8	-	-	61.4	10.3	18.3	78
	14	0.22	15.6	-	-	0.5	-	-	71.6	12.6	15.8	66
	15	0.19	19.6	-	-	0.01	-	-	91.6	8.4	-	1.0003
	16	0.10	20.8	-	-	6.7	-	-	75	-	25	84
	17	0.17	22.6	-	-	0.01	-	-	98.1	1.9	-	1.0003
	18	0.11	29.7	-	-	4.8	-	-	100	-	-	1.0003
	19	0.15	32.2	-	-	3.2	-	-	100	-	-	1.0003
Convent steel	20	0.04	1.2	0.02	0.008	-	-	18.3	8.8	100	-	1.02
	21	0.002	0.50	0.08	0.010	0.035	0.045	-	-	-	100 α	900

Table 2

Steel type	Thickness	Tensile Test			Volume fractions of the phase after tensile tests(%)		
		Yield Strength (kg/mm ²)	Tensile Strength (kg/mm ²)	elongation (%)	γ (austenite)	ϵ -martensite	α' -martensite
Steel of the invention	1	0.8	24.5	54.8	50.0	100	-
	2	-	19.7	50.4	57.4	100	-
	3	-	22.8	56.8	67.7	100	-
	4	-	26.3	58.2	61.2	100	-
	5	-	19.9	53.8	48.8	100	-
	6	-	19.4	49.6	46.6	100	-
	7	-	24.7	55.2	43.5	100	-
	8	-	18.6	58.5	58.6	100	-
	9	-	22.8	65.4	59.6	100	-
	10	-	19.0	50.4	52.8	100	-
	11	-	20.6	50.7	42.4	100	-
	12	-	26.4	55.7	43.9	100	-
Comparative steel	13	-	21.8	66.1	20.4	48.8	25.9
	14	-	29.0	83.8	14.0	44.1	13.7
	15	-	32.2	91.7	19.7	81.1	18.9
	16	-	25.5	51.5	37.0	52.4	-
	17	-	26.1	82.4	29.1	65.8	34.2
	18	-	21.5	53.0	37.2	100	-
	19	-	19.0	46.0	36.8	100	-
	20	-	23.5	65.5	79.2	80	-
	21	-	19	38	42	-	100 α
Convent steel							20

As shown in Table 1 above, the steels 1-12 of the present invention did not form ϵ -martensites and α' -martensites, but only formed austenite phase, so that they should be non-magnetic steels.

5 Meanwhile, the comparative steels 13-17 which departs from the composition of the steel of the present invention in their manganese and aluminum formed α' -martensites to have magnetic properties, and or formed ϵ -martensites.

10 The conventional steel 20 and the comparative steels 18 and 19, which have larger amounts in manganese and aluminum compared with the composition of the present invention had austenitic single phase, and had no magnetic property. The conventional steel 21 which is usually extra low carbon steel had a ferrite phase (α), and had magnetic properties.

15 On the other hand, in the case of the comparative steels 13-15 and 17, their tensile strength was high, but their elongation was very low. This is due to the fact that the contents of manganese and aluminum were too low, thereby producing ϵ -martensites and α' -martensites through a strain-induced transformation.

20 The comparative steel 16 showed a low elongation, and this is due to the fact that the content of aluminum was too high (although the content of manganese was relatively low), thereby forming α' -martensites through a strain-induced transformation, with lack of twins.

25 The comparative steels 18-19 showed low tensile strength and low elongation, and this is due to the fact that manganese and aluminum were too much added, resulting in that there was produced no martensite through strain-induced transformation, as well as no twins.

Meanwhile, the conventional steel 20 which is the normal stainless steel showed a high tensile strength and a high elongation. However, it had magnetic properties due to the formation of α' -martensites through a strain-induced transformation. Meanwhile, the conventional steel 21 which is a extra low carbon steel showed a tensile strength markedly lower than that of the steel 1-12 of the present invention, and this is due to the fact that the conventional steel 21 has a ferrite phase.

10

<Example 2>

On the steels 2 and 9 of the present invention, on the comparative steels 14 and 18, and on the conventional steel 21 of Example 1, formability limit diagram tests 15 were carried out, and the test results are shown in Figure 2.

As shown in Figure 2, the steels 2 and 9 of the present invention showed a superior formability compared with the conventional extra low carbon steel 21, because 20 twins were formed in the former. The comparative steels 14 and 18 shows no acceptable formability because they did not form twins.

Meanwhile, as shown in Table 2, the steels 1-12 of the present invention, which meet the composition range of 25 the present invention, showed a yield of 19-26 kg/mm², a tensile strength of 50-70 kg/mm², and a elongation of 40-68%. Particularly, the high elongation of the steels 1-12 of the present invention owes to the formation of twins through the tensile deformation. This fact can be 30 confirmed by the electron micrograph of the steel 5 of the present invention as shown in Figure 3.

In Figure 3, the white portion indicates twins, while the black portions (Matrix) indicate the austenite.

<Example 3>

5 A steel having the composition of Table 3 was melted under vacuum, and then, ingots of 30 kg were prepared from it. Then a solution treatment was carried out, and then, a slab rolling was carried out to form slabs of a thickness of 25 mm. This slab was heated to 1200°C, and a hot 10 rolling was carried out, with the finish rolling temperature being 900°C, thereby producing hot rolled sheets of a thickness of 2.5 mm. A microstructure observation was carried out on the hot rolled sheets to measure the size of the austenite grains, and the results 15 of these test are as shown in Table 3-A below.

Then the hot rolled sheets were subjected to measurements of yield strength, tensile strength and elongation. After such tests, a uniformly elongated portion of the tensile specimen after the tensile test was 20 cut out to subject to an X-ray diffraction test, thereby measuring the volume fractions of the phases. The result of this test is shown in Table 3-A below.

Table 3

Steel type	Chemical Composition (weight %)					
	C	Mn	Al	P	S	Ti
Steel of the invention	22	0.64	15.5	3.0	-	-
	23	0.38	17.9	3.3	-	-
	24	0.27	19.1	3.2	-	-
	25	0.47	23.1	3.5	-	-
	26	0.07	23.8	1.1	-	-
	27	1.43	25.1	0.8	-	-
	28	0.13	25.3	0.3	-	-
	29	0.98	28.5	6.0	-	-
	30	0.43	28.7	0.5	-	-
Comparative steel	31	1.12	34.7	2.5	-	-
	32	0.06	14.4	2.8	-	-
	33	0.19	19.6	0.01	-	-
	34	0.10	20.8	6.7	-	-
	35	0.17	22.6	0.02	-	-
	36	1.60	33.1	1.7	-	-
Convent. steel	37	0.60	37.0	3.3	-	-
	38	0.002	0.50	0.035	0.08	0.010
						0.045

Table 3-A

Steel Sheet No.	Thickness (mm)	Austenite Grain Size (μm)	Tensile Test			Volume fractions of the phases after the tensile tests			Steel Type of the Invention 22
			Yield Strength (kg/mm ²)	Tensile Strength (kg/mm ²)	Elongation (%)	γ (austenite)	ϵ -martensite	α' -martensite	
22	2.5	34	26.4	56.2	50.7	100	-	-	
23	"	35	21.2	54.4	54.6	"	-	-	23
24	"	29	25.6	59.9	61.8	"	-	-	24
25	"	30	30.1	69.6	41.5	"	-	-	25
26	"	30	21.3	61.5	55.8	"	-	-	26
27	"	32	33.4	70.1	40.3	"	-	-	27
28	"	35	21.9	54.6	50.7	"	-	-	28
29	"	30	28.1	63.2	40.8	"	-	-	29
30	"	33	27.4	57.0	43.1	"	-	-	30
31	"	34	29.7	67.4	41.7	"	-	-	31
32	"	32	24.5	68.5	18.8	52.9	23.7	24.4	Comparative Steel 32
33	"	30	33.1	91.5	17.6	85.3	14.7	-	" 33
34	"	31	27.1	54.7	37.4	100	-	-	" 34
35	"	34	28.4	69.2	27.5	90.6	9.4	-	" 35
36	"	30	33.4	72.2	26.4	100	-	-	" 36
37	"	35	28.6	58.5	34.4	100	-	-	" 37

As shown in Table 3-A above, the hot rolled steel sheets 22-31 which were manufactured according to the composition range and the hot rolling conditions of the present invention showed superior properties. That is, 5 they showed a tensile strength of 54-70 kg/mm², and a elongation of over 40%, and this owes to the fact that deformation twins were formed as a result of tensile deformation.

After the tensile tests, the steels 22-31 all showed 10 an austenitic single phase, and the lattice structure of the deformation twins was of face centered cubic structure corresponding to that of the austenite phase, with the result that they cannot be distinguished through an X-ray diffraction test.

On the other hand, in the case of the hot rolled comparative steels 32, 33 and 35, the tensile strength showed high, but the elongation was low. This is due to the fact that the contents of manganese and aluminum were too low, resulting in that ϵ -martensites and α' -martensites were formed through a strain-induced transformation. 15 20

The comparative hot rolled steels 34 and 37 showed a low tensile strength and a low elongation, and this is due to the fact that the contents of manganese and aluminum were too high, so that not only the formation of martensite through a strain-induced transformation could not occur, but also twins could not be formed. 25

Meanwhile, the comparative hot rolled sheet 36 showed a high yield strength and a high tensile strength, but a low elongation, and this is due to the fact that the content of the carbon was to high so as for carbides to be 30

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precipitated too much.

Further, the hot rolled steel sheets were cold rolled to a thickness of 0.8 mm, and this cold rolled steel sheets were annealed at a temperature of 1000°C for 15 minutes. Then on each of the test pieces, a microstructure observation was carried out to measure the austenite grain size. Then tensile tests were carried out to measure yield strength, tensile strength and elongation. Further, a uniformly elongated portion of the tensile specimen after the tensile tests was cut out to subject it to an X-ray diffraction test. In this way, the volume fractions of the phases was measured, and the result of the measurements are shown in Table 3-B below.

Further, the steel 24 of the present invention as listed in Table 3-B was observed by an electron microscope, the result of the observation being shown in Figure 4.

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Table 3-B

Steel Type	Thickness (mm)	Austenite Grain Size after annealing (μm)	Tensile test			Volume Fractions of the Phases			Remarks
			Yield Strength (kg/mm ²)	Tensile Strength (kg/mm ²)	elongation (%)	γ (austenite)	ε -martensite	α' -martensite	
22	0.8	35	24.5	54.8	50.0	100	-	-	22
23	"	38	19.7	50.4	57.4	"	-	-	23
24	"	34	22.8	56.8	67.7	"	-	-	24
25	"	31	29.7	68.2	43.5	"	-	-	25
26	"	37	18.6	58.5	58.6	"	-	-	26
27	"	39	31.3	70.4	41.0	"	-	-	27
28	"	36	19.0	50.4	52.8	"	-	-	28
29	"	35	27.6	60.7	42.4	"	-	-	29
30	"	36	26.4	55.7	43.9	"	-	-	30
31	"	38	26.8	65.4	44.6	"	-	-	31
32	"	32	21.8	66.1	20.4	48.8	25.9	25.3	32
33	"	36	32.2	91.7	19.7	81.1	18.9	-	33
34	"	34	25.5	51.5	37.0	100	-	-	34
35	"	36	26.1	68.4	29.1	90.8	9.2	-	35
36	"	35	31.5	71.5	27.2	100	-	-	36
37	"	38	27.2	56.0	36.8	100	-	-	37
38	"	65	19	38	42	-	-	100	-

As shown in Table 3-B above, the steels 22-31 of the present invention which meet the composition of the present invention had a tensile strength of 50-70 kg/mm² which is almost twice that of the conventional steel 38 which had a tensile strength of 38 kg/mm². Meanwhile, the elongation of the steels 22-31 showed to be over 40%, while the phase after the tensile tests showed to be an austenitic single phase.

On the other hand, the comparative steels 32, 33 and 10 35 showed a high tensile strength but a low elongation. This is due to the fact that the contents of manganese and aluminum were too low, resulting in that ϵ -martensites and α' -martensites were formed through a strain-induced transformation.

Meanwhile, the comparative steels 34 and 37 were low 15 in both the tensile strength and in the elongation, and this is due to the fact that the contents of manganese and aluminum were too high, so that no martensite phase through a strain-induced transformation as well as twins could not be formed.

Meanwhile, the comparative steel 36 was high in its yield strength and tensile strength, but low in its elongation, and this is due to the fact that the content of carbon was too high so as to precipitate too much carbides.

Meanwhile, the conventional steel 38 which is a extra 30 low carbon steel showed its tensile strength to be markedly lower than that of the steels of the present invention, and this is due to the fact that the steel 38 had a ferrite structure.

As described above, the steels 22-31 of the present

invention which meet the composition of the present invention showed a yield strength of 19-31 kg/mm², a tensile strength of 50-77 kg/mm², and a elongation of 40-68%. Particularly, the high elongation of the steels 22-31 of the present invention owes to the formation of twins through the tensile deformation. This fact can be confirmed by the electron micrograph for the steel 24 of the present invention as shown in Figure 4.

In Figure 4, the white portion indicates twins, while the block portion indicates the austenite structure (matrix).

<Example 4>

The formability limit tests were carried out on the steels 23 and 26, the comparative steel 35 and the conventional steel 38 of Example 3, and the result of the tests is shown in Figure 5.

As shown in Figure 5, the steels 23 and 26 showed the formability to be superior to that of the conventional steel 38 which is a extra low carbon steel, while the comparative steel 35 showed the formability worse than that of the conventional steel 38. This is due to the fact that, while the steels 23 and 26 of the present invention have a superior formability owing to the formation of twins, the comparative steel 35 forms ϵ -martensites, thereby aggravating the formability.

<Example 5>

A steel having the composition of Table 4 below was melted, and ingots of 30 kg were prepared from it. Then a solution treatment was carried out, and then, a slab

rolling was carried out into slabs of a thickness of 25 mm.

Here in Table 4, the steels 39-40 of the present invention and the comparative steels 54-60 were melted in vacuum, while the comparative steel 61 and the steels 50-53 containing a large amount of nitrogen (N) were melted under the ordinary atmosphere.

The slab which was prepared in the above described manner was heated to a temperature of 1200°C, and was hot-rolled under a finish temperature of 900°C to produce hot rolled steel sheets of a thickness of 2.5 mm. These hot rolled steel sheets were subjected to a microstructure inspection, thereby measuring the size of the austenite grains. The result of this inspection is shown in Table 4-A below.

Further, the hot rolled steel sheets were subjected to tensile tests to decide yield strength, tensile strength and elongation. After carrying out the tensile tests, the uniformly elongated portion of the tensile specimen was cut out to subject it to an X-ray diffraction test, thereby estimating the volume fractions of the phases. The results of these tests are shown in Table 4-A below.

Table 4

(Unit: weight%)

	Composition Steel type	C	Si	Mn	Al	Cr	Ni	Cu	Nb	V	Ti	N
Steel of the invention	39	0.13	-	16.1	5.5	-	3.9	-	-	-	-	0.005
	40	0.94	-	19.7	3.7	7.2	-	-	-	-	-	0.005
	41	0.44	-	20.3	5.6	-	-	-	0.2	0.4	-	0.006
	42	0.35	-	22.5	1.8	-	-	-	0.3	-	0.07	0.009
	43	0.08	-	24.6	3.6	-	-	-	-	0.3	0.14	0.009
	44	1.18	0.16	27.4	1.5	-	-	-	-	-	0.15	0.009
	45	1.35	-	27.8	2.2	-	-	2.7	-	-	-	0.006
	46	0.37	-	29.5	3.3	1.2	1.4	-	0.1	-	-	0.007
	47	0.28	-	32.3	2.1	-	-	0.4	0.1	-	-	0.006
	48	0.63	0.08	32.8	0.34	-	-	-	-	-	-	0.006
	49	0.13	0.22	33.5	1.2	-	-	2.8	-	-	-	0.005
	50	0.53	0.05	26.4	3.7	-	-	-	-	-	-	0.19
	51	0.45	0.05	27.4	1.2	-	-	-	-	-	-	0.09
	52	0.35	0.07	25.0	1.2	-	-	-	-	-	-	0.08
	53	0.40	0.20	26.5	2.3	-	-	-	-	-	-	0.10
Comparative steel	54	0.12	-	16.1	2.7	10.2	-	-	-	0.07	0.09	0.006
	55	0.13	-	19.3	1.4	-	-	-	-	0.61	0.44	0.007
	56	0.16	-	24.4	5.4	-	4.6	-	-	-	0.51	0.007
	57	0.24	-	27.4	4.7	-	0.4	-	1.3	-	-	0.006
	58	0.13	0.16	30.1	0.3	-	-	6.4	-	-	-	0.006
	59	0.75	0.35	32.9	3.3	1.8	-	2.5	1.1	-	-	0.008
	60	1.27	0.97	36.6	5.2	0.5	-	-	-	-	-	0.006
	61	0.44	0.05	27.2	2.3	-	-	-	-	-	-	0.23

Table 4-A

Steel Sheet No.	Thickness (mm)	Austenite Grain Size (μm)	Tensile Test			Volume fractions of the phases			Remarks (Steel Type)	
			Yield Strength (kg/mm ²)	Tensile Strength (kg/mm ²)	Elongation (%)	γ (Austenite)	ϵ -Martensite	α' -Martensite		
Hot rolled steel sheet of the invention	39	2.5	32	27.2	63.4	43.5	100	-	-	Steel of the Invention 39
	40	-	35	26.4	63.0	44.7	-	-	-	40
	41	-	34	21.8	61.1	40.4	-	-	-	41
	42	-	32	28.7	66.4	43.9	-	-	-	42
	43	-	31	25.4	63.6	44.2	-	-	-	43
	44	-	33	24.9	69.8	58.8	-	-	-	44
	45	-	35	23.3	60.2	40.2	-	-	-	45
	46	-	29	25.1	60.6	42.7	-	-	-	46
	47	-	34	23.2	60.8	44.4	-	-	-	47
	48	-	30	24.7	61.5	40.8	-	-	-	48
	49	-	33	26.2	60.4	49.6	-	-	-	49
	50	-	35	28.7	67.7	43.7	-	-	-	50
	51	-	31	28.9	63.5	45.4	-	-	-	51
	52	-	30	27.4	63.0	46.0	-	-	-	52
	53	-	34	29.3	66.7	46.5	-	-	-	53
Comparative hot rolled steel sheet	54	-	35	33.1	90.7	15.4	89	-	11	Comparative Steel 54
	55	-	34	27.5	68.3	17.9	100	-	-	55
	56	-	32	25.6	64.5	29.5	100	-	-	56
	57	-	32	24.7	61.5	25.8	100	-	-	57
	58	-	31	23.4	60.8	35.3	100	-	-	58
	59	-	30	21.6	62.9	30.7	100	-	-	59
	60	-	36	20.7	63.4	28.2	100	-	-	60
	61	-	34	26.8	69.7	25.5	100	-	-	61

As shown in Table 4-A, the hot rolled steel sheets 39-53 of the present invention showed a yield strength of 22-30 kg/mm², a tensile strength of 60-70 kg/mm², and a elongation of 40-60 %.

5 Further, the hot rolled steel sheets 39-53 of the present invention had fine austenite grain sizes down to 40 μm, while they do not form ε-martensites and α'-martensites even after undergoing the tensile deformation, but holds fully austenite phase. The reason why the steels
10 39-53 of the present invention showed such a high elongation of over 40% is that twins were formed during the tensile deformation.

Of the steels of the present invention, the hot rolled steel sheets 39-46 and 48-53, in which large amounts
15 of solid solution hardening elements such as Cr, Ni, Cu, Nb, V, Ti, N and the like were added, showed yield strengths and tensile strengths higher than those of the hot rolled steel sheet 47 of the present invention in which the solid solution hardening elements were added in smaller amounts.
20 This is due to the fact that the addition of the solid solution hardening elements results in the increase of the strengths.

Further, of the steels of the present invention, the hot rolled steel sheets 50-53 of the present invention, in which nitrogen was added in a large amount, showed higher yield strengths and higher tensile strengths over those of the hot rolled steel sheets 39-49 in which nitrogen was added in a smaller amount. This is due to the fact that fine twins are formed during the deformation caused by the
25 aluminum nitrides which were formed in the solidification stage, during the hot rolling stage and during the
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annealing heat treatment after the cold rolling.

Meanwhile, the comparative hot rolled steel sheets 58 and 60, in which Cu and Si were added in larger amounts over the composition of the present invention, showed an austenitic single phase, but their elongation is too low. This is due to the fact that non-metallic impurities and cracks formed during the rolling contributed to lowering the elongation.

Further, the comparative hot rolled steel sheets 55-57 and 59 in which Nb, V and Ti were added in amounts larger than the composition range of the present invention showed a low elongation, and this is due to the fact that the carbides were produced in large amounts within the steel to lower the elongation.

The comparative hot rolled steel sheet 54 which contained Cr in an amount larger than the composition range of the present invention showed high strengths, but its elongation was too low. This is due to the fact that a large amount of α' -martensites are formed after the tensile deformation.

The comparative hot rolled steel sheet 61 in which nitrogen (N) was contained in an amount larger than the composition range of the present invention showed a low elongation; and this may be due to the fact that nitrides were too much precipitated.

The hot rolled steel sheets which had been manufactured in the above described manner were cold-rolled to a thickness of 0.8 mm, and then, were annealed at a temperature of 1000°C for 15 minutes. Then a microscopic structure observation was carried out to decide the size of the austenite grains, and then, the tensile tests such

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as yield strength, tensile strength and elongation were carried out. Then the uniformly elongated portion of the tensile specimen after the tensile test was cut out to decide the volume fractions of the phases, and then, a 5 cupping test was carried out using a punch of a 33 mm diameter to measure the limit drawing ratio (LDR). The results of these tests are shown in Table 4-B below.

In Table 4-B below, the value of LDR is defined to be $LDR = [\text{diameter of blank}]/[\text{diameter of punch}]$. The 10 standard LDR for automobile steel sheets in which a good formability is required is known to be 1.94. Resorting to this standard, the formability were evaluated based on whether a steel sheet has an LDR value over or below 1.94.

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Table 4-B

Steel Type	Thickness (mm)	Austenite Grain Size after annealing (μm)	Tensile test			Formability test LDR*	Volume Fractions of the Phase			Remarks	
			Yield Strength (kg/mm ²)	Tensile Strength (kg/mm ²)	elongation (%)		τ (Austenite)	ε -Martensite	α' -Martensite		
Steel of the invention	39	0.8	34	26.3	63.2	42.4	1.94	100	-	-	39
	40	-	39	24.9	61.8	43.5	-	100	-	-	40
	41	-	37	20.6	59.7	40.6	-	100	-	-	41
	42	-	32	27.2	64.6	45.0	-	100	-	-	42
	43	-	35	24.7	60.2	45.6	-	100	-	-	43
	44	-	34	23.0	65.2	61.7	-	100	-	-	44
	45	-	37	22.0	58.4	40.6	-	100	-	-	45
	46	-	33	22.7	58.8	43.5	-	100	-	-	46
	47	-	38	21.2	57.7	45.9	-	100	-	-	47
	48	-	34	23.3	59.3	42.4	-	100	-	-	48
	49	-	36	26.4	58.2	48.8	-	100	-	-	49
	50	-	37	26.5	65.7	44.0	-	100	-	-	50
	51	-	33	26.2	61.1	44.2	-	100	-	-	51
	52	-	33	25.7	60.5	46.9	-	100	-	-	52
	53	-	35	25.9	63.3	47.1	-	100	-	-	53
Comparative steel	54	-	35	32.7	91.3	14.0	1.94 or less	87	-	13	54
	55	-	36	26.1	67.8	19.7	-	100	-	-	55
	56	-	32	24.3	62.8	30.4	-	100	-	-	56
	57	-	36	24.2	60.7	27.5	-	100	-	-	57
	58	-	34	22.6	58.6	37.1	-	100	-	-	58
	59	-	35	20.8	62.8	31.8	-	100	-	-	59
	60	-	39	19.4	61.3	28.6	-	100	-	100	60
	60	-	36	26.4	67.6	27.5	-	100	-	100	61

* LDR value = $\frac{\text{Diameter of blank}}{\text{Diameter of punch}}$

As shown in Table 4-B, the steels 39-53 of the present invention showed a yield strength of 20-27 kg/mm², a tensile strength of 57-66 kg/mm², and a elongation of 40-60%.

5 Further, the steels 39-49 of the present invention did not form ϵ -martensites or α' -martensites, but showed an austenitic single phase structure, thereby forming a highly stable steel. Further, they had a elongation of over 40%, and also showed superior formability. This owes 10 to the fact that twins are formed during the tensile deformation.

Among the steels of the present invention, the steels 39-46 and 48-53, in which the solid solution hardening elements such as Cr, Ni, Cu, Nb, V, Ti N and the like were 15 added in large amounts, showed high yield strength and tensile strength over the steel 47 of the present invention in which the solid solution hardening elements were added in smaller amounts. This owes to the fact that the solid solution hardening elements resulted in the increase of the 20 strengths.

Further, among the steels of the present invention, the steels 50-53, in which nitrogen was added in large amounts, showed higher yield strength and tensile strength over the steels 39-49 of the present invention in which nitrogen was added in smaller amounts. This owes to the 25 fact that nitrides were precipitated in reaction with Al in the solidification stage, during the hot rolling stage and during the annealing heat treatment after the cold rolling, and that fine twins were formed during the deformation caused by the aluminum nitrides.

Meanwhile, the comparative steels 58 and 60 in which

Cu and Si were added in excess of the composition range of the present invention showed an austenitic single phase, but their formability was not acceptable. This is due to the fact that the formability is aggravated by non-metallic impurities and fine cracks formed during the rolling.

Further, the comparative steels 55-57 and 59 in which Nb, V and Ti were added in excess of the composition range of the present invention showed an unacceptable formability. This is due to the fact that the carbides produced within the steel lowered the formability.

The comparative steel 54 in which Cr was added in excess of the composition range of the present invention showed high strengths, but low elongation and formability. This is due to the fact that a large amount of α'' -martensites were formed after the tensile deformation.

The comparative steel 61 in which nitrogen (N) was added in excess of the composition range of the present invention showed aggravated elongation and formability, and this is due to the fact that the nitrides were precipitated excessively.

<Example 6>

The steel 44 of the present invention as shown in Table 4 of example 5 was hot-rolled and cold-rolled in the same way as in Example 5. Then the cold rolled steel sheet was annealed under the annealing condition of Table 5 below.

After carrying out the annealing, a microstructure inspection was carried out on the cold rolled steel sheets, and then, tensile tests were carried out to decide the yield strength, tensile strength and elongation. A

cupping test using a punch of a 33 mm diameter was carried out to decide the formability, the result of these tests being shown in Table 5 below.

Table 5

Steel Number	Annealing Conditions		Austenite Grain Size After Annealing (μm)	Tensile Test			Forma- bility LDR value	
	Annealing Temp.	Annealing period		Yield Strength (kg/mm ²)	Tensile Strength (kg/mm ²)	elong- ation (%)		
Steel of the invention	62	600°C	20 sec.	4	58.9	87.6	41.9	2.06
			1 min.	4	56.1	86.8	42.8	2.06
			20 hrs.	6	48.4	82.9	48.7	2.06
	63	800°C	20 sec.	10	40.8	77.7	53.5	2.06
			1 min.	10	40.0	78.9	51.5	2.06
			20 hrs.	15	39.9	78.4	51.7	2.06
	64	900°C	20 sec.	19	39.2	74.3	54.9	2.06
			1 min.	20	38.0	73.5	55.1	2.06
			20 hrs.	24	34.9	70.6	57.2	2.06
Comparative steel	65	1000°C	20 sec.	31	23.7	65.6	60.0	1.94
			1 min.	30	23.1	64.4	61.2	1.94
			20 hrs.	34	23.0	65.2	61.7	19.4
	66	520°C	15 min.	-	97.9	106.6	11.7	1.94 or less
			30 hrs.	-	95.2	107.2	8.2	-
	67	800°C	4 sec.	-	94.4	107.8	7.4	-
			30 hrs.	28	24.2	67.3	32.8	-
	68	1050°C	20 sec.	53	20.1	56.2	57.7	-
			1 min.	53	20.4	57.0	50.4	-
			20 hrs.	57	21.8	56.4	53.6	-

As shown in Table 5, the steels 62-65 of the present invention which meet the annealing condition and the composition of the present invention have characteristics such that the austenite grain size after the annealing was

reduced to below 40 μm , that the yield strength, the tensile strength and the elongation were high, and that the formability is superior.

On the other hand, the comparative steels 66-68, which meet the composition of the present invention, but which depart from the annealing conditions of the present invention, have the following characteristics. That is, in the case where the annealing temperature was lower than the annealing temperature range of the present invention, or where the annealing time was short, the austenitic structure was not recrystallized so as to give high strengths, but the elongation and the formability were too low. On the other hand, in the case where the annealing temperature was too high or where the annealing time was too long, the austenite grains was coarsened so as for the elongation to be bettered, but the formability was aggravated due to the formation of carbides within the steel.

<Example 7>

The steel 44 of the present invention and the conventional steel 38 as shown in Table 4 of Example 5 were hot-rolled and cold-rolled in the manner of Example 6, and then, an annealing was carried out at a temperature of 1000°C for 15 minutes.

Then, on the annealed steel sheets, a spot welding was carried out with the condition of: a pressure of 300 kgf, a welding current of 10 KA, and a current conducting time of 30 cycles (60 Hz). Then hardness tests were carried out on the welded portion at the intervals of 0.1 mm with a weight of 100 g, the result of this test being

illustrated in Figure 6.

As shown in Figure 6, the weld metal, the heat affected zone and the base metal of the steel 44 of the present invention showed a vickers hardness value of 250 5 in all the three parts, and this is an evidence to the fact that the steel 44 of the present invention has a superior weldability.

The reason why the steel 44 of the present invention has such a superior weldability is that there is generated 10 no brittle structure layer on the heat affected zone.

On the other hand, the conventional steel 38 showed that the weld metal and the heat affected zone had a vickers hardness value of about 500 which is much higher than the base material. This is an evidence to the fact 15 that its weldability is an acceptable, brittle phases being formed on the weld metal and the heat affected zone.

According to the present invention as described above, the steel of the present invention has a tensile strength of 50-70 kg/mm² which is twice that of the extra low carbon 20 steel. Therefore, the weight of the automobile can be reduced, and the safety of the automobile can also be upgraded. Further, the solubility limit is very high, and therefore, the carbon content can be increased to 1.5 weight %, so that no special treatment is needed, and 25 that a special management for increasing the formability is not required in the process of cold rolling. Consequently, an austenitic high manganese steel having superior formability, strengths and weldability can be manufactured.

What is claimed is:

1. An austenitic high manganese steel having superior formability, strengths and weldability, its composition comprising:

5 less than 0.70 weight % of carbon;

Mn and Al being added without departing from the enclosure formed by A, B, C, D and E as shown in attached Figure 1; and

10 the balance including Fe and other indispensable impurities.

2. The austenitic high manganese steel having superior formability, strengths and weldability as claimed in claim 1, wherein the formability is improved by 15 deformation twins.

3. An austenitic high manganese steel having superior formability, strengths and weldability, the steel comprising:

20 a composition of less than 1.5% of C, 15.0-35.0% of Mn, and 0.1-6.0% of Al, in weight %, the balance consisting of Fe and other indispensable impurities. and the size of austenite grains being less than 40.0 μm .

25 4. The austenitic high manganese steel having superior formability, strengths and weldability as claimed in claim 3, wherein the formability is improved by deformation twins.

30 5. An austenitic high manganese steel having superior formability, strengths and weldability,

the steel comprising:

a composition of (in weight %) less than 1.5% of C,
15.0-35.0% of Mn, and 0.1-6.0% of Al,

one or more elements being added by selecting them
5 from the group consisting of: less than 0.60% of Si, less
than 5.0% of Cu, less than 1.0% of Nb, less than 0.5% of
V, less than 0.5% of Ti, less than 9.0% of Cr, less than
4.0% of Ni, and less than 0.2% of N, and

10 the balance consisting of Fe and other indispensable
impurities, and

the size of austenite grains being less than 40.0 μm .

6. The austenitic high manganese steel having
superior formability, strengths and weldability as claimed
15 in claim 5, wherein the formability is improved by
deformation twins.

7. A process for manufacturing an austenitic high
manganese steel having superior formability, strengths and
20 weldability, comprising the steps of:

preparing a steel slab with a composition of (in
weight %) less than 1.5% of C, 15.0-35.0% of Mn, and 0.1-
6.0% of Al, the balance consisting of Fe and other
indispensable impurities, and

25 hot-rolling said slab in the known method.

8. The process for manufacturing an austenitic high
manganese steel having superior formability, strengths and
weldability as claimed in claim 4, wherein the hot
30 rolling is carried out at a temperature of 1100-1250°C,
with the finish temperature being 700-1000°C.

9. A process for manufacturing an austenitic high manganese steel having superior formability, strengths and weldability, comprising the steps of:

5 preparing a steel slab, said steel slab having the composition of (in weight %) less than 1.5% of C, 15.0-35.0% of Mn, 0.1-6.0% of Al, and one or more elements selected from a group consisting of less than 0.60% of Si, less than 5.0% of Cu, less than 1.0% of Nb, less than 0.5% of V, less than 0.5% of Ti, less than 9.0% of Cr, less than 4.0% of Ni, and less than 0.2% of N, the balance including Fe and other indispensable impurities, and
10 hot-rolling said steel slab in the known method.

15 10. The process for manufacturing an austenitic high manganese steel having superior formability, strengths and weldability as claimed in claim 6, wherein the hot rolling is carried out at a temperature of 1100-1250°C, with the finish temperature being 700-1000°C.

20 11. A process for manufacturing high manganese steel having superior formability, strengths and weldability, comprising the steps of:

25 preparing a steel slab having the composition of (in weight %) less than 1.5% of C, 15.0-35.0% of Mn, 0.1-6.0% of Al, and the balance of Fe and other indispensable impurities,

hot-rolling said steel slab, and cold-rolling the hot-rolled steel sheet in the known method, and

30 annealing the cold-rolled steel sheet at a temperature of 500-1000°C for 5 seconds to 20 hours.

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12. A process for manufacturing an austenitic high manganese steel having superior formability, strengths and weldability, comprising the steps of:

5 preparing a steel slab, said steel slab having the composition of (in weight %) less than 1.5% of C, 15.0-35.0% of Mn, 0.1-6.0% of Al, and one or more elements selected from a group consisting of less than 0.60% of Si, less than 5.0% of Cu, less than 1.0% of Nb, less than 0.5% of V, less than 0.5% of Ti, less than 9.0% of Cr, less than 10 4.0% of Ni, and less than 0.2% of N, the balance including Fe and other indispensable impurities,

hot-rolling said steel slab, and cold-rolling the hot- rolled steel sheet in the known method, and

annealing the cold-rolled steel sheet at a temperature 15 of 550-1000°C for 5 seconds to 20 hours.

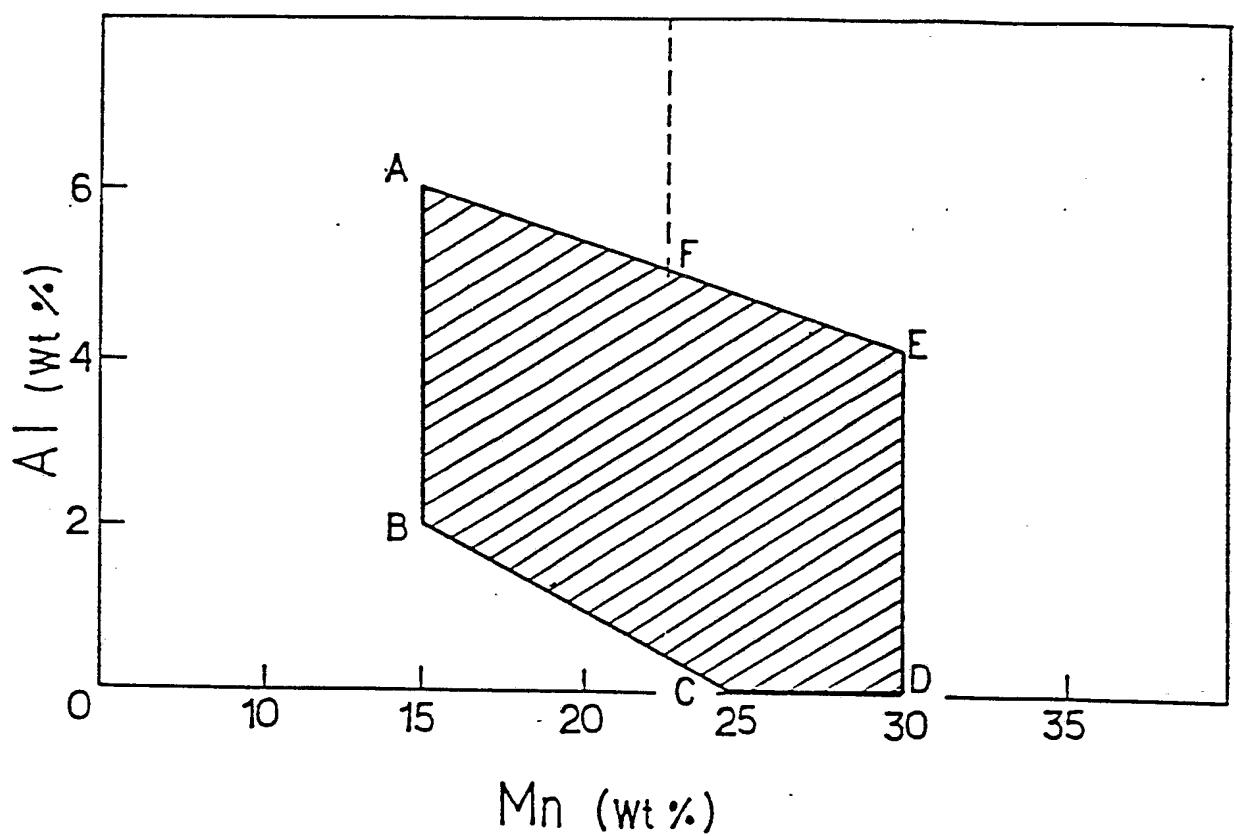
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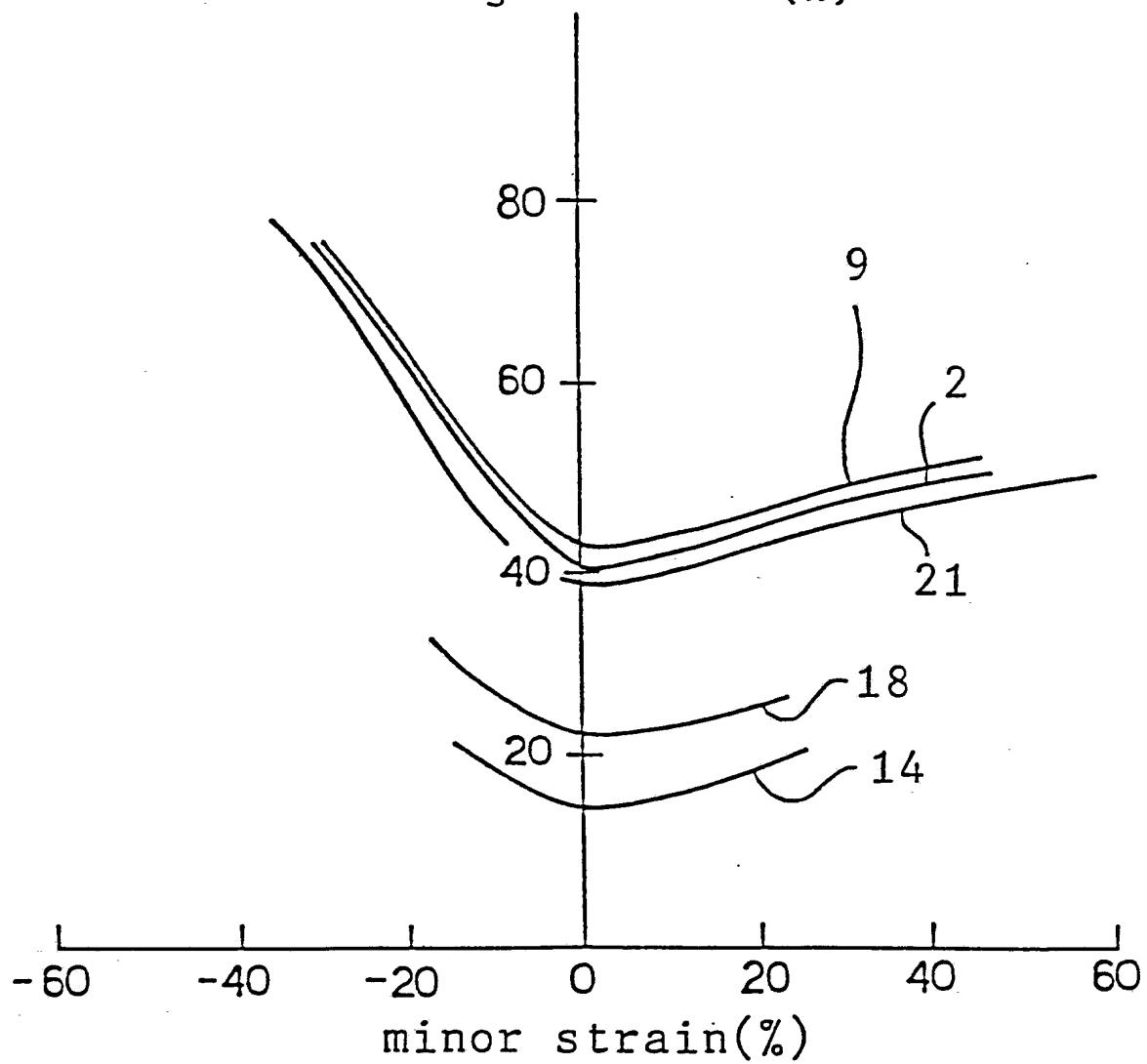
FIG. I



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FIG.2

major strain(%)

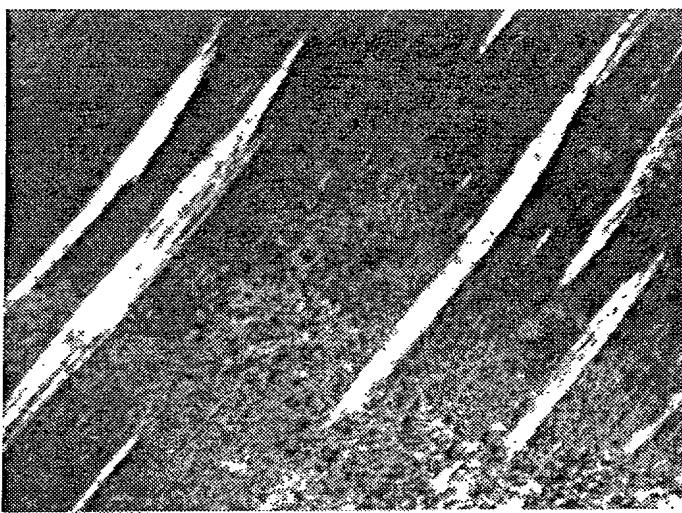


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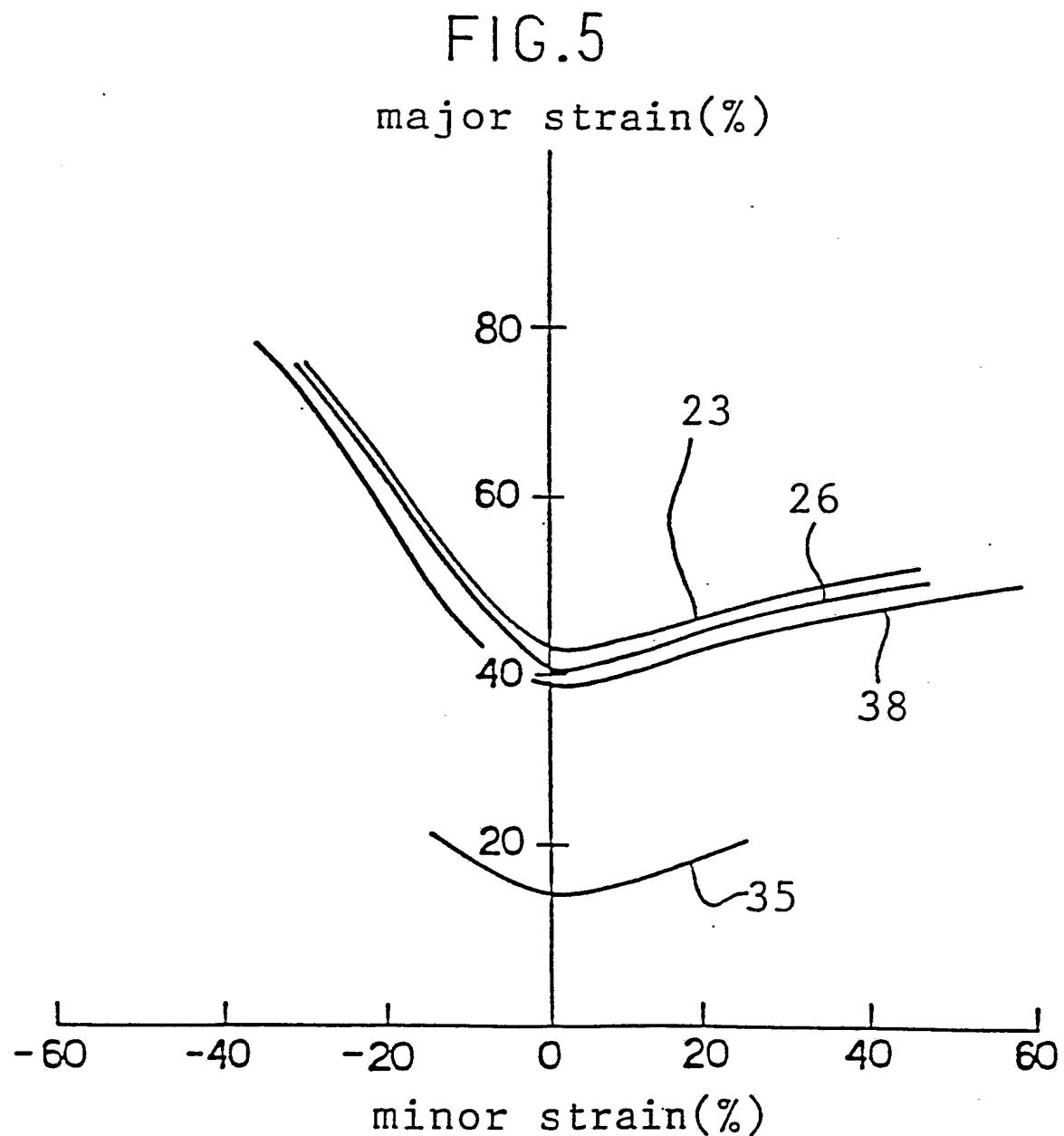
FIG.3



FIG.4

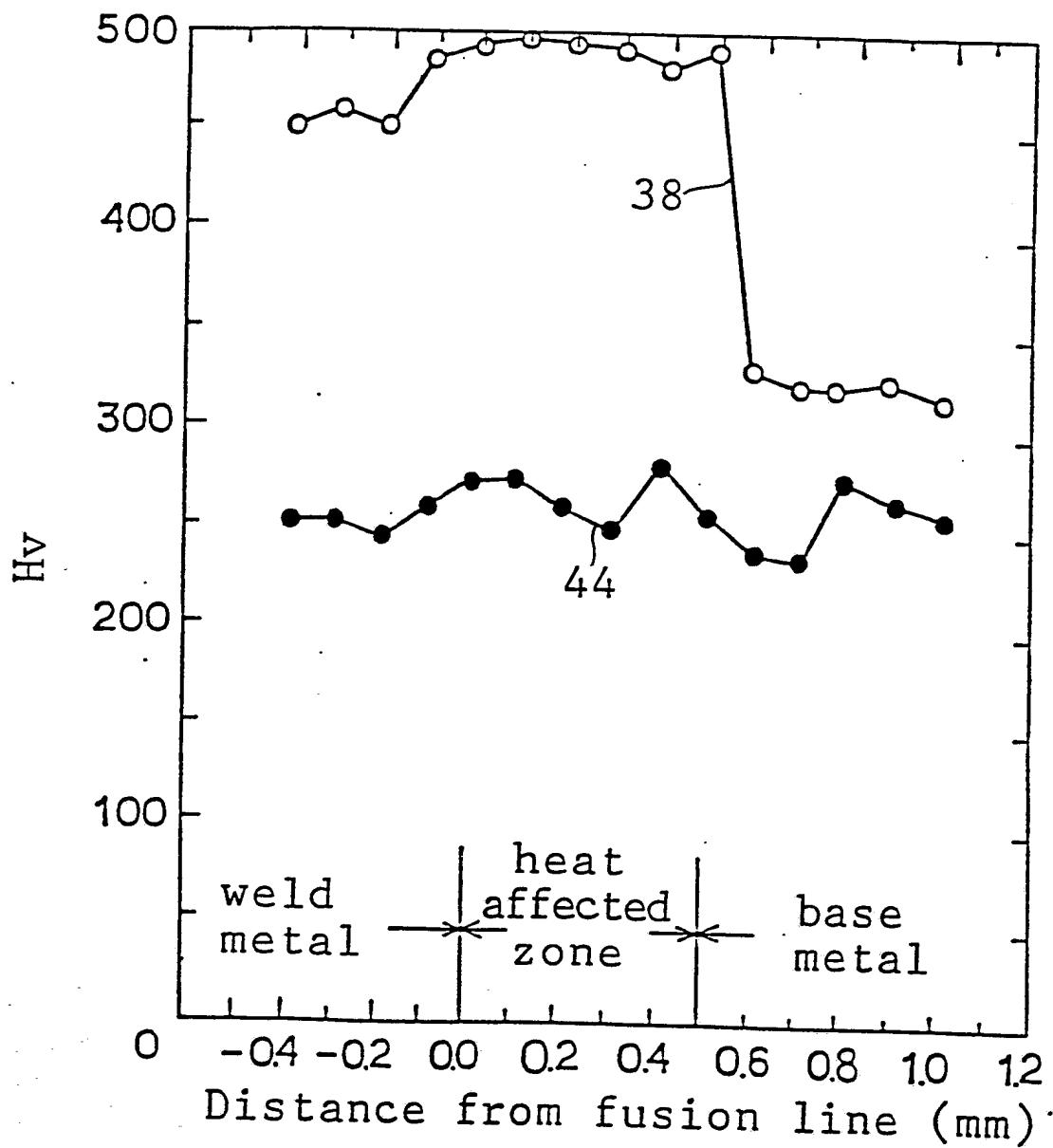


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FIG.6



INTERNATIONAL SEARCH REPORT

International application No.

PCT/KR 92/00082

A. CLASSIFICATION OF SUBJECT MATTER

IPC⁵: C 22 C 38/06, 38/08, 38/12, 38/14, 38/16, 38/18; C 21 D 7/13, 7/02
 According to International Patent Classification (IPC) or to both national classification and IPC

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)

IPC⁵: C 22 C 38/04, 38/06, 38/08, 38/12, 38/14, 38/16, 38/18, C 21 D 7/13, 7/02

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

AT; Chemical Abstracts

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	US, A, 4 847 046 (KIM et al.) 11 July 1989 (11.07.89), claims.	1-12
A	EP, A1, 0 414 949 (IPSCO) 06 March 1991 (06.03.91), claims 1,3.	1,7
A	DE, A1, 3 903 774 (NATIONAL SCIENCE) 16 August 1990 (16.08.90).	1,7
A	Chemical Abstracts, Volume 105, no. 26, issued 1986, December 29, (Columbus, Ohio, USA), Adler, P. H. et al. "Strain hardening of Hadfield manganese steel", page 234, the abstract no. 230 599t, Metall. Trans. A. 1986, 17A(10) 1725-37. -----	2,4,6

<input type="checkbox"/>	Further documents are listed in the continuation of Box C.	<input checked="" type="checkbox"/> See patent family annex.
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* Special categories of cited documents: "A" document defining the general state of the art which is not considered to be of particular relevance "E" earlier document but published on or after the international filing date "L" document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified) "O" document referring to an oral disclosure, use, exhibition or other means "P" document published prior to the international filing date but later than the priority date claimed	"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention "X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone "Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more others such documents, such combination being obvious to a person skilled in the art "&" document member of the same patent family
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Date of the actual completion of the international search 24 March 1993 (24.03.93)	Date of mailing of the international search report 30 March 1993 (30.03.93)
Name and mailing address of the ISA/ AT AUSTRIAN PATENT OFFICE Kohlmarkt 8-10 A-1014 Vienna Facsimile No. 0222/53424/535	Authorized officer Lux e.h. Telephone No. 0222/5337058/31

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.

PCT/KR 92/00082

In Recherchenbericht angeführtes Patentdokument Patent document cited in search report	Datum der Veröffentlichung Publication date	Mitglied(er) der Patentfamilie Patent family member(s)	Datum der Veröffentlichung Publication date
Document de brevet cité dans le rapport de recherche	Date de publication	Membre(s) de la famille de brevets	Date de publication
US A 4847046			keine - none - rien
EP A1 414949	06-03-91	US A 4944814 WO A1 9103579 AU A1 42161/89 BR A 8907902	31-07-90 21-03-91 08-04-91 01-09-92
DE A1 3903774	16-08-90	US A 4968357	06-11-90